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15.4. Damage production and accumulation (Session Organizers: M. Kiritani and N.M. Ghoniem)

Damage production and accumulation

1. M. Kiritani (Chairman)

Before we start this session, I would like to have a few words on how we will manage it. In ICFRM-4 in Kyoto, we had a discussion session on cascade characteristics and defect production. In this session, we plan to proceed to damage production and accumulation. In Kyoto, which was very similar to this one in size, we had nine short prepared presentations followed by 73 questions, answers and comments, and we must manage the same size this time.

I observed other discussion sessions yesterday. Some of them were like series of paper presentations, but I would rather like more discussions from the audience. After a keynote presentation, we have here ten panelists. Each panelist will be given five minutes for short stimulating presentations, followed by discussions from the audience. At the ending part, Nasr Ghoniem will summarize the session by 10 min.

2. Keynote presentation by M. Kiritani (Hiroshima Institute of Technology, Japan)

I will attempt to present what we know now, particularly in recent years from an experimental perspective in this area of research. First I would like to emphasize that there is a vast difference between thin foil and bulk irradiation results for studies of cascade damage. The majority of clusters (~96%) in thin foil irradiations of D–T neutrons of nickel are of the vacancy type, while the microstructure of a comparable bulk irradiation case is almost entirely of the interstitial loop type. Subsequent high-energy electron irradiation of these samples shows that the small clusters shrink, indicating that they are of the vacancy type. Larger clusters that grow under electron irradiation are determined to be of the interstitial type.

The three-dimensional structure of sub-cascades can be determined from stereoscopic analysis of irradiated samples. The size of sub-cascades and the distance between their centers can also be immediately obtained from this method. If we consider high Z materials, we see shorter distances between sub-cascades, while if we consider copper, nickel and steel, the distance is almost twice as large.

One of the most reliable methods for obtaining the size of the collision zone is by using ordered alloys to examine the size of the disordered region. There is a wide range of recoil energies in neutron irradiations. However, we can always estimate the average energy deposition density and the size of cascades as functions of the Primary Recoil Energy (PKA). The apparent energy density reaches about 10 eV per atom, and that may correspond to temperature increases of about 40 000 K, in good agreement with MD simulations. If we do similar types of analysis, in which we plot the number of subcascades versus the recoil energy for different materials, we observe large differences between different materials. If we do hundreds of different computer simulations with a wide distribution of PKA energies, we would be able to get reasonable statistics from computer simulations, which can be compared to experiments. However, we see little work in this area because it requires a large computational effort.

When we raise the temperature, there are several different structures of cascades. This is due to intracascade reactions, by both diffusional and dynamic processes. At high temperatures, collisions produce only large clusters, while at lower temperatures, these cascades produce smaller size clusters. The knowledge of freely migrating defects can be evaluated for a number of different types of experiments. This is a thin foil irradiation. Near the surface, there is many more vacancy clusters than in its inside. If we analyze the distribution of sizes from the surface to the inside of the foil, we can determine the number of interstitials that had reactions inside the samples.

Another example of arrival rates is the nucleation and growth process of clusters. As far as nucleation is concerned, we seem not to understand accurately how it occurs. If we observe pre-existing dislocations (i.e. before irradiation), their climb during irradiation would be a good indication of the arrival rate at the dislocation.

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Thus, we can estimate the fraction of freely migrating interstitials. Related to dislocations, however, there are many topics that are very important. One of them is the enhancement of interstitial cluster nucleation near dislocations, but I will not get into this topic now. It is a combined subject of nucleation from cascades (i.e. heterogeneous) and from freely migrating defects. Their growth, however, is primarily due to freely migrating defects. If we talk about interstitial cluster nucleation, the number density of interstitial loops depends on many variables, such as the material, temperature and so on, but it is always less than the number density of collision cascades. The density of this vacancy type of STF in nickel thin foils is almost exactly the same as the density of collision cascades. However, in bulk materials, the density of interstitial loops is almost two orders of magnitude smaller than cascade density. Computer simulations show that the number of sub-cascade clusters per cascade can be as high as 10. However, experimental observations show that this number can be as small as 0.01 in bulk materials. Still the number of interstitial loops that succeed to grow is even less than that. It may be on the order of 0.0001 of the original density of cascades. This is a current theoretical problem to handle at the present time. From these types of experiments, we can determine the fraction of freely migrating defects which escape the cascade zone, how many end up in clusters, dislocations, grain boundaries, etc.

If an interstitial cluster contains only five or six interstitials, we know that it moves. However, I believe that even larger clusters containing up to 100–200 interstitials are still able to migrate. Crowdion motion occurs along the $\langle 1\ 1\ 0\rangle$ direction in FCC and the $\langle 1\ 1\ 1\rangle$ direction in BCC metals. The one-dimensional motion of crowdions appears to have a very large effect on the microstructure under neutron irradiation. For example, in one grain we find a lot of dislocations, while in a neighboring grain we find only the dot structure of cascades. In nickel, because the $\langle 1\ 1\ 0\rangle$ direction intersects the surface, all small loops escape to the surface. Such effects can be explained if we accept the one-dimensional motion of interstitial loops. When we add electron irradiation to neutron irradiated specimens, about 80% of the loops which have Burgers vectors which can glide parallel to the surface remain within the sample, while those with Burgers vectors pointing to the surface disappear. This was a very puzzling result at first, because neutron irradiation should have no orientation dependence. However, the one-dimensional motion concept can now explain these results.

The last part of my presentation will cover damage accumulation. If we examine many types of irradiations, we can systematically show four stages of behavior. As can be seen from the data, there is a 'super-linear regime', followed by a linear regime for single cascades. The third regime is square root relationship with the fluence because of partial annihilation. Finally, we reach 'saturation' in the fourth regime, where we have full geometric overlap of clusters. We observe some basic differences between fission and fusion neutron effects on the density of clusters, in some cases by factors up to 20.

3. Prepared discussion by N.M. Ghoniem (UCLA, USA)

My presentation is intended to stimulate discussions rather than summarize information on damage production and accumulation mechanisms, because this has been already covered in my invited talk earlier. I would like to elicit discussions on a number of questions, which appear to be still unresolved in a definitive way. The two suggested broad topics of discussion are the following:

1. Dynamic nucleation of vacancy and interstitial clusters in cascades.
2. Production versus absorption bias.

In the area of dynamic nucleation of defect clusters in cascades, we are basically interested in events which are not driven by diffusion, but which occur on a very short time scale during cascade evolution. For FCC metals, the conditions for the collapse of vacancy loops and the formation of stacking fault tetrahedra (STFs) need to be clarified. In other words, we still need to answer the question of whether cascade collapse occurs within the lifetime of one single cascade, or it still requires the cooperative effects of other neighboring cascades where energy is exchanged? Presently, Professor Kiritani points out to the possibility of intra-cascade effects as important in determining the final configuration of the cluster. On the other hand, computer simulations by Dr. de la Rubia show that a 'zone refinement' process, in which a re-solidification front propagates rapidly from the cascade center, drives the collapse mechanism. In BCC metals, we do not see SFT's and vacancy loops, while they are readily observed in FCC. Such behavior must be related to the elastic constants or differences in stacking fault energies. A clear answer has yet to be made.

Now I will turn my attention to the damage accumulation question. As we saw earlier from Professor Kiritani's presentation, we may be able to divide up the range of fluence into four different regimes. On a log–log scale, we can schematically show that the first regime represents the emergence of invisible clusters that are constantly being 'compacted' by the impact of neighboring cascades. We may call this first stage: the 'super-linear' regime. At some low dose, we get to the second regime, where the fraction of zones to be compacted becomes very small, and each new cascade produces a corresponding cluster. This is followed by the third regime, where partial annihilation may take place, as a result of the interaction between vacancy and interstitial

clusters. It is typically observed in this regime that the density of clusters is proportional to the square root of the fluence. Finally, in the fourth regime, we arrive at full geometric overlap between different cascade contents leading to a saturation of the number density with fluence. The main mechanistic factors, which control the transition between those regimes, are not entirely determined. For example, we may ask whether the transition from regime 2 to 3 is related to the length of the Replacement Collision Sequence (RCS), or to the geometric size of the cascade itself.

Computer simulations suggest that interstitial clusters nucleate at the periphery of cascades. However, we still need to know the driving force for their nucleation at the cascade periphery. If it is their tendency to reduce the overall elastic strain energy, then how is it achieved during such short time scales, and what is the role of elastic anisotropy in their formation? It is also shown now by computer simulations, as well as indirect experimental evidence that small interstitial clusters migrate one-dimensionally in Crowdion configurations. Their directed migration is along the ⟨1 1 0⟩ direction in FCC and the ⟨1 1 1⟩ direction in BCC. However, the question of the size dependence of their migration still remains to be answered.

The final topic, which I would like to discuss here, is the topic of 'production' versus 'absorption' bias factor. We can clearly see now from MD simulations that within one cascade, there is an initial asymmetry in the production efficiency of both types of vacancy and interstitial clusters. However, subsequent events make this simple picture more complex. First, the one-dimensional diffusion of interstitial clusters will transport this initial asymmetry to distances that are far from the initial event. This leads to possible 'localization' of the vacancy rich core of the cascade, as suggested by Professor Kiritani in the 'CLIB' mechanism. In the third and fourth regimes of damage accumulation, we know that significant overlap occurs between individual cascades. In a given volume, this may have the tendency to reduce the initial asymmetry just by virtue of the statistics of interaction between neighboring cascades. In the early days of rate theory development, the absorption bias was found to explain well the swelling behavior of metals. However, now that we know the existence of a 'production bias', we need to clarify the phenomena which can be only explained by the 'production bias', those which can be explained only by the 'absorption bias', and those which require both.

3.1. Discussions

S. Ishino: I would like to comment on the first question raised by Dr Ghoniem. I think that it is almost clear that single cascades produce vacancy clusters. May be Professor Sekimura will shed more light on this question at a later point in our discussions. At very low dose, if we look at 100 000 times magnification, and at the 250–350 nm range, only few ions per second would be produced. Thus, we would be able to isolate vacancy clusters resulting from single cascades. So, I think that as both professors Kiritani and Ghoniem have shown, we can demonstrate, by our experiments, that the super-linear and linear regimes result from single cascades. The super-linear regime occurs only some times, and it appears to depend on the energy and type of PKA.

T.D. de la Rubia: Already six or seven years ago, we showed by MD simulations that single cascades can produce vacancy Frank loops, and we've been showing that over and over again in many metals. Well, that's true only in FCC metals. In particular, those metals of low melting point. In BCC metals, however, we do not find large vacancy clusters, as was shown by Bob Odette, Rodger Stoller, David Bacon so on and by us. I do not think that they are even observed experimentally, and may be Professor Kiritani can comment on that. I believe that the vacancy loop yield in BCC metals is very small. In FCC metals, it is also clear that they can be created by a single impact.

M. Kiritani: What is the main reason for the absence of vacancy cluster formation in BCC metals? Is it the low density of vacancies at the cascade center, or is it the intrinsic nature of BCC metals? Also, we know that even quenching and deformation do not result in vacancy clusters. Of course, we know that voids form at higher temperature. From my experience, I believe that in FCC crystals, such as aluminum, we never observed Stacking fault tetrahedra (SFTs). But this is wrong. If you give the proper conditions to aluminum, you can cover the entire sample with STFs. But in BCCs, the situation is different, even if we give the same treatment as aluminum to produce vacancy clusters and SFTs, you can never tell, you know. But the whole sample is still covered with distorted structures. I therefore cannot answer this question at this time, and I would say that the cascade collapse mechanism is still unknown in BCC metals.

B.N. Singh: Let me bring in a new element in something that has been discussed many years ago. I think that if we probably look at cascades produced by each event, counting each particle and determining the ratio of the number of clusters to the number of cascades, we can define an efficiency of vacancy loop formation. For copper, this efficiency is nearly 100%, as you go to nickel, you find that it is less efficient than copper. Gold and silver behave very much like copper. Now the situation in BCC indicates that they are very much inefficient in producing vacancy clusters. As Professor Kiritani said, the thin foil experiments are done such that all interstitials will come out of the sample, or go laterally. Thus, we mask the effects of interstitials. In these types of experiments, you are actually creating a

situation where the vacancy density within each cascade is very high, may be on the order of 100 vacancies or so. Of course, when you compare the results of experiments close to the surface, and inside the bulk, you will find more vacancy clusters near the surface. In fact, we can all calculate this effect, as Bob Averback did. From the linearity of the dependence of the density of tetrahedra on the dose, we can show that each cascade produces an SFT.

E. Kuramoto: I would like to make a short comment on the vacancy behavior in iron cascades. We made positron annihilation measurements at low temperatures for neutron-irradiated iron. We found that the positron lifetime in this case is not very long, indicating that it is almost the same as in a single vacancy. At these low temperatures, the concentration of vacancies is very high, but they seem to be all isolated with no significant clustering.

M. Kiritani: What if you did the same experiments for FCC at low temperatures? Do you observe only single vacancies as well?

E. Kuramoto: We did the same experiments on nickel. We can see that the positron lifetime is still similar to that of a single vacancy at the beginning, but then increases slightly indicating the existence of very small clusters. By isochronal annealing in FCC, we see a transition temperature for the formation of STFs.

M. Kiritani: This indicates that the starting point in FCC and BCC is the same, but it is only the diffusional phase that makes the difference.

T.D. de la Rubia: This is a very interesting point, because nickel and iron have about the same melting point, and they appear to behave the same, as far as cascade collapse is concerned. So, it is not the difference between the FCC and BCC structures, but the melting point, which determines the collapse efficiency. If the melting point is low, then you always get large vacancy clusters at the center of the cascade as the cascade cools. If the cascade is not very dense, and the melting point is high, then we never see vacancy loops at the cascade center. Therefore, the simulations and experiments are all consistent with the melting point idea.

M. Kiritani: I still do not understand this difference. Is it due to the crystal structure?

N.M. Ghoniem: There are still many factors that are not discussed in reference to cascade collapse. Does the elastic interaction play any role (and hence, elastic constants)? Also, what is the effect of one cascade on another one that has a 'loose' cluster of vacancies?

T.D. de la Rubia: David Bacon did most of the simulations on cascade overlap, and since he is not here, I cannot answer this question. I personally don't think that it does not play much of a role. As far as the elastic interaction is concerned, the energy per atom during the cool down phase is on the order of eV per atom, and I suspect that the elastic interaction energy is much smaller. I think that it is just a matter of whether you can sweep the vacancies to the center of the cascade or not.

4. Prepared discussion by H.L. Heinisch (PNL, USA)

I will talk about going beyond MD simulations and by that we talk about the modeling tools as well as the phenomena themselves. This is the end of the MD simulation, where nothing is happening from the MD standpoint (pico-second) time frame. However, a lot still has to happen beyond this phase. One of the ways to study cascades beyond this phase is the Kinetic Monte Carlo (KMC) Method, where you can go from say a pico-second to a nano-second. Now you've got rid of those interstitials that have moved away. You're looking at the cascade now in terms of the defects and not in terms of atoms, so you can learn some new things.

This is just an example where you can determine the fraction of defects that cluster, and those that escape the cascade and so on. Another thing you can do though is to go beyond one single cascade in this little region, and look at the interaction between cascades. You can look at defect accumulation by taking a larger volume and through the cascades in it, the cascades you get from Molecular Dynamics. You can do this kind of thing and look at cluster densities as the irradiation continues, and this is up to 0.1 dpa at a low dose rate. This can be compared with experiments. At the same time, you can vary all the parameters in it. Whatever you think can happen, you can put into the model.

We can determine that way quantities that cannot be directly measured from experiments. Things like surviving fraction of defects, invisible clusters, and the sensitivity to all these parameters. This is an old version of the timeline of cascade evolution, where we show the various phases during the cascade lifetime. It is also shown that there are different time and length scales associated with each physical phenomenon, and that we should consider them all together. For instance, to determine the fraction of Freely Migrating Defects (FMDs) at the end of one MD cascade is one thing, and to determine it after cascade annealing is another. Another issue in these models is the question of whether you can take MD cascades and put them together in Monte Carlo simulations. In this type of model, defects just jump around in a totally random and uncorrelated manner. Do we need to make the KMC more sophisticated? To bridge to rate theory is another step. Obviously, with atomic models, we cannot go up to 10 or 100 dpa. We need rate theory, which will tell us a lot of things that cannot be determined from KMC. So, the idea is that you're bridging from an atomistic picture to a more continuum picture.

4.1. Discussions

S.J. Zinkle: I would like to address your first question on the last viewgraph. In my mind, the appropriate definition of Freely Migrating Defects should be the number that escapes correlated recombination. Now that has created a lot of confusion in the literature, because some times people talk about FMDs but then include sink losses and so on. However, it should still go beyond MD, if there are additional in-cascade recombination taking place, after cascade cooling has finished. What I know people talk about is that there are probably not a lot of correlated recombination beyond this phase, so you're probably alright to first order in taking this fraction from MD calculations, but in principle you should take it to the next step.

H.L. Heinisch: I agree with Steve's comments. You are right in saying that there is not much additional recombination, but I do not know if anyone has tested that to see if it is an artifact of how the annealing is done at that stage. I do not think that it makes much of a difference.

M. Kiritani: In MD simulations, it is important to consider the details of cooling and physics of interaction, but we need statistical data. For 100 KeV recoils, we see wide variations in the results. But MD people say that only one calculation requires a lot of computing resources. So, I am asking if we can still get useful information on the statistics of these cascades.

T.D. de la Rubia: It is very easy to get statistics nowadays, because computers are much faster. Then you can get stochastic effects beyond MD by doing Kinetic Monte Carlo. As far as recombination out of the cascade, there are cases where significant recombination within the cascade takes place. These cases include in the Monte Carlo strain field effects, by including the capture radii of loops. These are not only one or two nearest neighbor distance, but much longer. For example, we've done that following the calculations of Professor Kuramoto. He calculated the extent of the dislocation bias in gold, and the extent of the interaction between an interstitial and a loop. When you put that in the simulation, you see that it is a strong effect, and thus has to be taken into account in KMC simulations.

M. Kiritani: In pioneering work, when you discover a new phenomenon, it is exciting, but to get useful statistics, it seems like a lot of routine work. Can you still do that? Are there many people who are going to do this?

C.H. Woo: About the statistics, we had an interesting conversation. We emphasize a lot of things around the average or mean value. In fact, a lot of things which have to do with evolution are those which are deviations from the mean. This is similar to the situation of the US population, for example, where only a small number of individuals are able to make the change. So, whether we do 10 examples out of 10^{23} events, what counts are the deviations in these statistics.

N.M. Ghoniem: I would like to add to this point about the role of computer simulations. We mention Kinetic Monte Carlo as a bridge between MD and rate theory, but I have a different opinion. I believe that KMC is a numerical technique for solving atom transport and diffusion problems in drift or random fields. It is not different from rate theory, but is a different method to bring about other features of the physics. The current trend to rely on computer generated data as the only means of understanding physics is a bit misleading, and we have to be cautious on the underlying physical assumptions.

T.D. de la Rubia: I do agree with Professor Ghoniem, but I believe that the number of approximations you do in KMC is much smaller than in chemical rate theory, where you're making approximations on homogeneous medium and mean field and so on. By virtue of its simplicity, and the lack of approximations as far as the possible kinetic paths, KMC may take us to places where rate theory did not.

N.M. Ghoniem: This assertion remains to be seen.

B.N. Singh: I go back to statistics. I think that I agree with Dr Woo that our main concern should be of a different type. It should be whether or not we understand the physics of computer simulations. Even if they give physical trends, which we can generally understand, then the statistics do not matter that much. As regards the comparison between KMC and rate theory, I think that Professor Ghoniem is quite right. I would even add more caution to that, because there is a general tendency today to say that if we can get away with computer experiments, let's forget about the real experiments. We have to keep going between theory and experiments in order to understand physics, otherwise we will lose track.

S.I. Golubov: The situation with Monte Carlo simulations and rate theory is almost symmetric. Because in rate theory, we have fluctuations in size space, but we lose fluctuation in real crystal space. In KMC, we can introduce fluctuations in real crystal space, because we have real cascades. But, here we lose the effects of fluctuations on size space, since we have to introduce periodic boundary conditions. We should use KMC in parallel with rate theory to improve our knowledge. At the end, we need to have a good theory and not a good computer.

5. Prepared discussion by B.N. Singh (Risø National Laboratory, Denmark)

Let me begin with simple experiments, where we show single crystal, high purity molybdenum irradiated with neutrons in the range of 10^{-4} – 0.2 dpa. For comparison, we have also, side by side, a molybdenum–

rhenium alloy single crystal. The thing to notice is that in a very short time, this single crystal is no longer single crystal. It has become completely ruined by rafts of interstitial loops all over the single crystal. If you look at clusters in-between, you can see the cluster and raft densities have strong variations across the crystal. Why does this happen in molybdenum, but does not happen in molybdenum–rhenium? We only have 5% Re, so why is such a big difference? This is only one example of interstitial loop rafts decorating dislocations, but you can find many other examples. I found one example of 316 stainless steel, where it was irradiated by nickel ions at 600°C, but you can still find decorated dislocations there. How does this happen and why?

I will now show you results of visible cluster densities as a function of dose for iron at 0.18 T_m and for nickel and copper for about 0.2–0.3 T_m. If you now compare the densities, you see quite a significant difference, and we must answer the question of why we have such a drastic difference. Continuing along the same lines, we compare Mo with Cu. This is the dependence of the experimental density of clusters on temperature. In copper, we have to get to stage V to see that the cluster density is going down. However, in Mo, as soon as we go beyond stage III, we get significant and steep drop in cluster density. Even when we put impurities in, it does not take us to the copper results. So, it confirms the points raised earlier as to the difference in BCC and FCC in general.

Now I would like to go to the problem of swelling, and this is work that we've done with J. Evans. This is DFR work taken at about 80 dpa. I normalized the swelling in FCC and BCC metals. It seems that in the case of BCC metals, there is no significant swelling, and no significant temperature dependence. The more important point to note here is that we are beginning to see swelling just below stage III. As vacancies become mobile in BCC, we form cavities. In FCC, however, you have to go to 0.35 T_m (i.e. beyond recovery stage V) before we can get cavity nucleation. So, there is this fundamental difference coming all the way from the primary defect structure to swelling, even at 80 dpa. So, whatever is happening, it is just not for the first 5 ps. It lasts for years. So, we have to think about these differences very hard. It is not something that we can wave our hands to, and say it is the fault of MD or TEM, but this is something real. I think that this is where I will leave, without any answer coming from side. As time goes along in discussions, I will try to answer some of those questions, but I would first offer those questions to the audience.

5.1. Discussions

M. Kiritani: I have a comment on your first slide, where you refer to molybdenum single crystal with alloying element additions. There may be several reasons for the stabilization of interstitial clusters. For example, it may be easier for them to nucleate at impurity sites, or that the one-dimensional easy motion is substantially suppressed. Under the microscope, we notice that when we add impurities, even small loops do not move. But, for pure metals, it is very easy even for large loops to migrate. This means that the addition of rhenium enhances nucleation, and gives the loops more of a chance to grow.

B.N. Singh: I agree with this assessment in principle. I would just add one more thing here. We are finding from both dynamic MD calculations, as well as static relaxation calculations that, in iron for example, when you form interstitial clusters, you cannot maintain them in sessile configuration. They very quickly transform to prismatic loops, and then they glide.

T.D. de la Rubia: I think that this is absolutely correct. In fact you can even generalize it, because in BCC metals in general, the stacking fault energy is very high. Therefore, your loops will always convert to perfect configurations. So, they cannot support a stacking fault. In FCC, the conversion probability will depend on the stacking fault energy. But, for FCCs with stacking fault energies below 60 ergs/cm^2, the loops may remain faulted, because they produced sessile in the first place. The question that remains to be answered is what is the effect of the local stress field in the cascade on loop unfaulting.

B.N. Singh: Brian Eyre, already 25 years ago, has introduced the idea that loop nucleation is controlled by the stacking fault energy. Therefore, this proposal to explain the difference between loop concentrations in BCC and FCC should be examined.

M. Kiritani: This idea of stacking fault energy controlling loop nucleation has been around for 20 years, and I do not accept it. The stacking fault energy, even the highest, is only a small fraction of point defect energy, and thus should not play a role.

T.D. de la Rubia: The stacking fault energy does not play a role in the nucleation. It controls whether the loop faults or unfaults. I am not saying that it controls whether we get a loop or not, I think that nucleation is much more controlled by the shear modulus of the material. This is a fundamental property of the material, and it does not come out of any simulation.

P. Vajda: I have just a naïve question regarding all those philosophical discussions between MD and rate theory and so on. Now that we've seen some experiments. Dr. Singh, why don't you then replace all the molybdenum by rhenium? This is a very nice HCP refractory, and no body talks about HCP metals around here. It would be interesting to see if it behaves like FCC or like a BCC metal with a high melting point.

B.N. Singh: If we did that, you will have the same problem as we are having in molybdenum, so you would

probably not be getting anything. I think that getting back to what Professor Kiritani was saying, the impurities will block the one-dimensional glide of those tiny interstitial clusters. You can see that the coalescence in molybdenum–rhenium results in a different kind of global growth and accumulation. But here, this kind of segregation of things is typical of one-dimensional motion.

S.J. Zinkle: I think that one other difference between the BCC and FCC is that in FCC you get vacancy clusters in the cascade, but in BCC you do not. This explains why you see right away the difference in the initiation of the swelling: stage III versus stage V. In FCC, you have to wait till stage V, because you do not have enough free vacancies. Stage V is when vacancy clusters begin to thermally dissociate, whereas in stage III isolated vacancies are not in the clusters, so you can get the swelling straight away.

C.H. Woo: This comment is about the HCP situation, because I happen to have worked a bit on zirconium. There are a lot of faulted loops in zirconium. In fact, all the c-type dislocation loops are all faulted.

6. Prepared discussion by S.I. Golubov (Institute of Physics and Power Engineering, Russia)

We have been discussing the properties of cascades for many years now, even at this conference where we heard a lot of material related to the behavior of cascades. Let me remind you of the title of our discussion session: it is damage production and accumulation. Someone may ask a naïve question, are they two different problems, or are they two sides of the same problem. I believe that they are two sides of the same problem, because they are strongly related. If we try to analyze the literature, we will find the meaning of cascade production. In electron irradiation, we have only Frenkel pair production, but under other conditions, we have the complications of cascade production. However, when people model damage production and accumulation, they often use standard rate theory, with the same production rate of vacancies and interstitials. We have now direct experimental evidence, which was shown earlier by Dr Singh, which shows many differences depending on the type of bombarding particle. So, we have information going from production of single Frenkel pairs to the production of large-size cascades.

In standard rate theory, the maximum swelling is associated with electron irradiation. In neutron irradiation, there are cascades that create for us another mechanism of defect accumulation, not present in the electron irradiation case. This mechanism is the production bias. Now what are the main features of cascade production theory? We have some set of parameters that give cascade efficiency, fraction of interstitials that go directly into interstitial clusters, and fraction of clustered vacancies. As was shown by Singh and Forman, if we use these two equations, we will have a catastrophic situation. This is because once you produce stable interstitial clusters, you cannot stop them. Their concentration will be so high for any small dose that they will kill radiation damage if any. Thus, we need another mechanism that can remove those interstitial clusters from the system. May be there are many ways to do this, but I will show you only one way which allows better agreement between theory and experiments. This equation describes for us the one-dimensional diffusion of interstitial clusters. You remember the experimental evidence presented by Dr. Singh for this, but there are many other experimental observations of one-dimensional glide of these small clusters. So, under neutron irradiation, we have three-dimensional diffusion of single vacancies, but one-dimensional diffusion of interstitial clusters.

6.1. Discussions

N.M. Ghoniem: I see that there is a high degree of emphasis on interstitial clustering, and the need to include them in rate theory is over-emphasized. However, since the early days of rate theory development, even if you go back to the work of Bullough, Eyre and Krishan, you will find that point defect clustering was included. It was recognized as a necessary condition for explaining differences between electron and neutron irradiation effects on swelling. Therefore, we have to put these issues in perspective because we tend to repeat history over again. There has been point defect clustering in the early rate theory models, and now we have some more modifications of the clustering concept by including interstitials as well.

S.I. Golubov: In the framework of the equations that I presented, you can see both ε_v and ε_i, for the fractions of clustered vacancies and interstitials, respectively. The ε_v is exactly the BEK model.

B.N. Singh: To say that we are re-discovering the wheel, you have to first recognize what the wheel is ? We clearly say that up to ε_v, the model is exactly BEK. By clearly stating that once you start introducing interstitial clustering and their removal, we have a different model from BEK. Please tell me what are we rediscovering. I would like to have the answer from you.

N.M. Ghoniem: My point is that when we put the story of rate theory together, we should recognize the initial fundamental idea of BEK. The success of explaining differences between neutron and electron irradiation swelling data relied on the existence of vacancy clustering at the cascade center. Now we are trying to see the additional clear features, which interstitial clustering would bring into these explanations, particularly for differences between BCC and FCC metals.

B.N. Singh: O.K. Let us take the spectrum effects in the results we've just shown. Please try to use just BEK, and I would be interested in how you would reproduce these results.

S.I. Golubov: It is necessary to emphasize that between interstitial clusters and vacancy clusters, we have a big asymmetry. Interstitial clusters are stable clusters, and one-dimensional diffusion is absolutely necessary.

C.H. Woo: Can I just say one word. I think that this issue has been reviewed in a keynote paper two years ago in the Hawaii conference. I think that there is no excuse for not having seen that paper yet.

7. Prepared discussion by H. Wollenberger (Hahn-Meitner-Institute, Germany)

Ladies and gentlemen, you might have realized already that the topics of atom transport, phase transformations and phase stability had only a very small fraction of time so far in this conference. This is despite the fact that we know that phase instabilities of structural materials has serious consequences on their properties. The reason why I am bringing up this topic here in connection with microstructure evolution is that there is really an important link between the two topics. This fact stems from recognition that atom transport is provided by freely migrating defects (FMDs). These are the defects that migrate randomly in a stochastic process, and they provide the means for atom transport. We have already heard the definition of FMDs, and here is one that has really practical consequences, because you can measure its effect directly in atom transport. This definition of FMDs does not agree exactly with that Dr Zinkle has given before, as those that escape correlated recombination. This definition came from the old time of electron experiments, where you had a very low density of defects, and then it was worthwhile to consider just those that escape correlated recombination. In the high-density cases, which we have these days, that is not such a good definition. If we consider the definition on the basis of atom transport, then it is very clear.

Any prediction of the atomic rearrangements, which we have under irradiation, requires knowledge of those FMDs. The adequate method to determine their concentrations is the self-diffusion coefficient. That is to say that we have to measure the self-diffusion coefficient under irradiation. For practical materials, you have non-negligible couplings between the fluxes of defects and solutes; the Kirkendall effect for example. In this case, you should try to measure FMDs by means of segregation. I will present just a brief examination of the types of self-diffusion measurements. The diffusion coefficient is given by a contribution of mixing the radiation enhanced one and the thermal diffusion. We are now interested in the radiation-enhanced diffusion. It is given by contributions from both vacancies and interstitials. Here we have the correlation factors, which couple the flux of defects with those for the atom transport. When you look closer to the steady state, and see the concentrations of defects, then you get the relationship which tells you that the defect production rate (K), and the fraction of FMDs divided by the diffusion coefficient and the sink strength (k^2). We have shown today in the poster session that as you separate these things from each other by careful diffusion measurements. When you are able to do that, and obtain the information on the sink strength on one side, which is all this clustering business you have heard before, you can get the fraction of FMDs. The experimental situation at the moment is that the depth resolution for such diffusion profiles is about 2 nm, and that means that you need a fluence of about 5 dpa for such an experiment. That is to say that you can do it with heavy ion irradiation to get it in the right time. You can also do it with neutron irradiation, but no one has done it up till now. It is very urgent to do these kinds of experiments. In principle, however, new methods are required to get much higher sensitivity. We believe that the with so-called tomographic atom probe, you can get depth resolution of 0.2 nm. Thus you can measure diffusion coefficients that are two orders of magnitude smaller than what we have now. Then, we could do such experiments with electron irradiation. I wanted to make a point in connection with what you have heard in all these discussions. We seem to rely mainly on the experimental results of electron microscopy. If it were the only method available, that would be very dangerous for long term prospects. You need to have other methods, and I think that atom transport methods are the ones that should be done much more than before.

7.1. Discussions

N.M. Ghoniem: One point of discussion is the definition of the fraction of freely migrating defects.

R.E. Stoller: What we measure by the segregation is the fraction of freely migrating defects. The measurements are accurate as long as they go but the influence you make from those measurement can be wrong if you don't think about it carefully, because you still lose freely migrating defects to sinks along the way. And so we have to be careful what we are calling about here.

H. Wollenberger: When you are doing segregation experiment you can get the fraction only. Of course you have to know the sink structure near the surface, no question.

S.J. Zinkle: Just to follow up the question by Stoller, why we do not want to use the definition that you and others established 30 years ago. That is simply the defect fraction relative to the NRT dpa of those escaping correlated recombination. That seems to me very clear

definition that everybody understands. Rate theory modelers can add the sink strength for dislocation loops, voids and whatever else, that the source term that goes in the model. You need defects that survive correlated recombination that are mobile.

H. Wollenberger: My point is very simple. When the defect concentration becomes too high, you can never decide between correlated or not correlated recombination. That is the point where the definition loses its sharpness.

B.N. Singh: We have to remember that the irradiation enhanced diffusion which have two components D_v and C_v, and in order to get that number you have to know the sink strength, otherwise you cannot get correct C_v. One of the problem is the couple of freely migrating defects to the value of sink strength, this is not easy to determine independently.

8. Prepared discussion by Y. Kato (Kyoto University, Japan)

I would like to talk about the defect accumulation and microstructural evolution at high doses, with an emphasis on the application of modeling work in this category to practical problems for fusion demonstration and power reactors. This issue should tightly be associated with the ongoing high performance material development activities rather than the very fundamental physics of radiation damage in pure metals and clean model alloys. A joint fusion reactor design team at National Institute for Fusion Science and Japanese universities is working on a helical type demonstration device, FFHR, in which the first wall/blanket structural materials will be subjected to neutron load of nearly 500 dpa over 30 years of its prospective lifetime. Therefore, this fluence level is our goal of radiation resistant material development.

For this purpose, the key understanding should be the fluence dependence of material behavior up to ultra-high doses. Other important issues, such as fusion neutron irradiation effect, fission–fusion correlation and varying condition effect at medium to high fluence levels are classified prerequisite understandings. The important facilities and methodologies for the above-stated understandings include 14 MeV intense neutron sources, fission reactors capable of material irradiation up to very high doses, material modeling and carefully designed modeling-oriented/supporting experiments.

For designing highly radiation resistant materials, commercial martensitic steels give us valuable suggestions. The 12% chromium martensitic steely endure fast neutron irradiation up to 200 dpa as proved so far. The reasons for superior radiation stability are strong sink condition, provided by high density dislocations, narrow strip martensitic lath structures, fine matrix precipitates and maybe other stable matrix clusters, and the dislocation decoration with alloying elements during irradiation. The former suppresses free defect flux and the latter weakens imbalance in defect partitioning.

Further minimization of defect flux should be possible by nano-particle dispersion, as demonstrated in MC-stabilized austenitic stainless steels, and the developmental ferritic ODS steels should be the test bench for this mechanism. The stability of nano-particles under irradiation will be a critical issue, therefore, the introduction of cascade defect clusters stable at the service temperature range should be attempted. Helium accumulation in fusion condition is another very critical phenomenon, since we cannot control the transmutation itself by any mean. However, it is possible to retain large amount of helium without precipitation in a certain microstructural condition. For example, in a nano-crystalline silicon carbide fiber, no helium precipitation was detected at helium concentration of 10 000 appm even after annealing at 1673 K as reported by Hasegawa, et al. Finally, I would like to emphasize that we already have witnessed several mechanisms of potential ultra-radiation resistance. The material for 500 dpa is not a dream with an appropriate material development strategy.

8.1. Discussions

N. Yoshida: As you said, important issue of high dose is helium effect. Although there are efforts such as the tritium trick, we need a high-energy neutron source.

B.N. Singh: At such a high dose of 450 dpa, I have to think beyond helium because many other impurities, phases, and precipitates ruin the grain boundary. We can change the whole chemistry. Even at 600°C, helium is rather slow diffusion species, and you can produce very fast moving impurities and they ruin grain boundary in no time.

N. Yoshida: Helium and defect interaction is so strong, and it modifies the defect structure strongly even at medium dose.

B.N. Singh: No, I do not agree. Helium is so inert, and interaction with vacancies is completely harmless. Change of materials in chemical way is far more dangerous.

A. Kohyama: When we think about such a high dpa, another important issue is nuclear transmutation. It is not the continuing issuers of accumulating dpa, but it is a matter of material design. For high dpa we have to start from redesign of materials.

9. Prepared discussion by N. Yoshida (Kyushu University, Japan)

(Reproduced from the written resubmission from the author.) In the case of fusion experimental devices and

fusion reactors in future, the temperature of the plasma facing components may change widely due to the repetition of discharges. However in the most of the irradiation experiments carried out so far for fusion materials development, the irradiation temperature was tried to fix during the irradiation. Here, I should like to emphasize that the variation of irradiation environment, especially variation of irradiation temperature, plays essential roles for defects accumulation and therefore performance of materials under the fusion environment.

In the recent studies it was reported that temperature variation affected strongly on microstructural evolution such as formation of interstitial loops, voids and precipitates. In the case of Fe–16Cr–17Ni alloy, for example, accumulation of interstitial loops and voids are suppressed by step irradiation at 200/400°C and temperature cycle irradiation at 200/400°C and 300/450°C at JMTR. In this case the highly accumulated vacancies by the irradiation at the lower temperatures, where the thermal mobility of vacancies is low, become mobile at the higher temperature and mutually annihilate by reacting with interstitial loops formed at the lower temperature. In the case of Fe–16Cr–17Ni–0.25Ti alloy, however, the accumulation of voids was unexpectedly enhanced by the temperature cycle irradiation at JMTR; void swelling/dpa reaches about 105/dpa at 0.1 dpa. In the case of Fe–16Cr–17Ni–0.1P alloy, on the other hand, formation of Fe_2P precipitates is modified by the temperature variation irradiation and as a result suppression of void swelling by the precipitation is drastically reduced.

As mentioned above cyclic temperature variation and step temperature variation may change microstructural evolution very much even if at high dose, and the effects depend on the many factors such as materials, variation temperature range, dose and so on. Studies in this field have just started. We need systematic irradiation under temperature variation and theoretical works to understand these essential phenomena for evaluation of materials performance under the fusion environment.

9.1. Discussions

H. Matsui: I want to emphasize the impact of temperature variation. In binary V-based alloys, microstructure evaluation is very much accelerated by temperature change especially when the two temperatures are close to each other. The nucleus formed at lower temperature survive at higher temperature, and evolve to dense microstructure. When the lower temperature is very low, the nucleus formed there become unstable at higher temperature and they do not have direct effect.

B.N. Singh: Quite some years ago, in DFR they did in situ experiment of temperature change of 100 degrees, the swelling rate went up by factors of 10 and so the creep rate. Lot of things happen by the temperature change.

10. Prepared Discussion by M. Victoria (PSI, Switzerland)

My comment may be a little bit out-of-order, since we have not reached really the question of mechanical properties, but I refer to Dr Singh's lecture in which, among the impacts of the accumulation of the defect microstructure, he mentioned, at low to medium temperature regime, after deformation you produce these dislocation channeling. What is happening is the dislocation sweeps through the obstacles created by irradiation and destroys the obstacles on the slip plane and that leads to the formation of clear channeled regions because other dislocations than immediately be triggered in the same slip plane. It is shown by us in single crystals in my laboratory and other places, and you can calculate at least 2% shear strain associated with the development of such channeling. You can ask yourself the question what happens when the band reaches grain boundary, whether it is possible for the grain boundary to relax such a strain. We have now some idea of what happens in iron. What you are looking at now is the surface of iron tensile micro-specimens, 300 μm thick and the grain size of about 30 μm. In the unirradiated specimen what you see of the necking region essentially small microcrack which is starting to develop and propagate in the necking region. Now as you look at the surface of the irradiated specimen, what you see is this. There is the distribution more or less uniform in the surface of this small objects, when you look at with a higher magnification you will see that they are actually grain boundary cracks, and they are clearly associated with this dislocation band formation in the crystal. We have seen the same story in irradiated 304. Another final test of the story, this is the case in 316 stainless steel which deforms at room temperature by twining. When the irradiated crystal deforms by twining, these dislocation channeling is arrested and you do not see any crack at room temperature. If you go to 300°C and deform it there is no more deformation twining and you see the cracks again. I think we have whole ranges of implications that go from the kind of information that we are trying to get out of micro-specimens to irradiation associates stress corrosion cracking, where I think it plays the roll of these cracking mechanism of the right temperature and doses.

10.1. Discussions

M. Kiritani: Have not you observed the inside structure of crack at the grain boundary? It corresponds to the end of the channel?

M. Victoria: It corresponds to the end of the channel.

M. Kiritani: I think that in the next ICFRM, it is not a clean subject, but some discussion session microstructures are OK but concentrating on high dose irradiation otherwise the mechanism low dose cascade and high dose mixed then problem will be diverged, so we must suggest next organizer to have a specific discussion on high dose irradiation and mechanical properties.

R.E. Stoller: I have some additional questions and comments on the presentations given by Drs Singh and Golubov regarding the comparison between experiment and theory for comparison of neutron, electron and proton irradiations. Although the point defect and defect cluster model is very detailed, it is not clear how void formation is simulated. This is particularly important for such a comparison of different irradiation environments. Since the dose is so low in this experiment, and the maximum swelling is only $10^{-2}\%$, the results would be very sensitive to the details of void formation. Both the proton and neutron irradiations involve gas (H, He) as a potential agent that accelerates nucleation. This is lacking in the electron irradiation case. Furthermore, cascade residue (even from protons) could play a role in promoting void formation. This is also lacking in electron irradiations. Hence, the low swelling in the electron case could be a result of different void nucleation at the lower displacement rate. Previous electron irradiations that showed easier void formation were performed at 1000 times higher displacement rate. If the experiments could be carried out to higher doses and swellings, I am not convinced that the current ranking would still persist.

P. Vajda: This comment is directed to N.M. Ghoniem. When calculating the RCS lengths, you only consider the length reduction due to an increase in temperature. You should, however, also take into account the significant decrease in the threshold energy with an increase of temperature. In copper, the displacement energy decreases from $\sim$20 eV at 4K to $\sim$10 eV at 300 K. This might play a role in the calculation of the number of escaping single interstitials at the cascade periphery, as observed by electron microscopy.

Jinnan Yu: Professor Ghoniem, the physics of cascade production in metals is very clear. However, in semiconductors and insulators, there exist strong ionization effects. These processes may influence the displacement threshold energy and cascade processes. What do you think about this problem?

N.M. Ghoniem: In general, you may be correct in your observations regarding the sketchy state of knowledge on cascade physics in semiconductors and insulators. However, there have been computer simulations of collision cascades in silicon, and in silicon carbide that I am aware of. In silicon, de la Rubia's computer simulations show melting in THE cascade core, followed by amorphization as the solidification front propagates.

B.N. Singh: I have a number of remarks and comments on the origin of the production bias and on defect clusters. (a) In addition to the asymmetry in the production of defect clusters, the difference in the thermal stability between clusters of SIAs and vacancies is the second major physical feature, which leads to the production bias. (b) Clusters of SIAs and their properties in FCC and BCC crystals: (I) SIA clusters in BCC Fe are configurationaly unstable in sessile form. In other words, practically all SIA clusters produced in BCC Fe are glissile. This is not the case in FCC Cu. (ii) Clustering efficiency of SIAs is higher in FCC Cu than that in BCC Fe.

11. Session summary by N.M. Ghoniem (UCLA, USA)

Professor Ghoniem gave an overall summary of the session discussions and presentations. The summary closely follows the individual discussions led by discussion leaders, as follows.

11.1. Experimental observations

Many features of defect production and accumulation can be obtained by careful experimental observations, and by employing the proper combination of methods. Professor Kiritani (Hiroshima University) showed a variety of experimental methods, which may be summarized as:

1. The character of a dislocation loop (vacancy or interstitial type) can be identified by subsequent electron irradiation. The growth or shrinkage of the dislocation loop can be directly related to its character.
2. The size of the collision cascade can be determined from experimental observations of disordering of intermetallic ordered alloys, such as Cu_3Au.
3. TEM observations are used to reveal the dependence of the number of sub-cascades on the energy of the Primary recoil Atom (PKA).
4. It is experimentally observed that by increasing the sample temperature during irradiation, groups of sub-cascade defects can agglomerate, resulting in a reduction in the number of sub-cascades and an increase in their size.
5. Several experimental possibilities are available for determination of the number of Freely Migrating Defects (FMD). These are: (a) experiments with wedged samples, where the variation in the size of cascade and loop microstructure can be used to determine surface proximity effects on FMDs. (b) TEM observations of direct dislocation climb have a direct correlation with the FMD.
6. In situ experimental observations are useful in verification of the one-dimensional glide of dislocation

loops. Gliding dislocation loops containing several hundred atoms have been experimentally observed.

11.2. Stochastic annealing and kinetic Monte Carlo studies

Dr Heinisch (PNL, USA) summarized the state of research on studies of large cascades with the Monte Carlo method. In these studies, defects produced within the cascade zone are allowed to jump and 'anneal'. However, the physics of annealing (e.g. recombination distance) is calibrated with MD computer codes. Good agreement with TEM observations of cascade size, and distribution of sub-cascades has been demonstrated. This method, termed Kinetic Monte Carlo (KMC) appears to be suitable for volumes containing one or few cascades. The proper role and expected range of validity of the KMC and Rate Theory methods for description of microstructure evolution have not been clearly established by the discussion participants.

11.3. Defect accumulation

Dr Singh (Risø, Denmark) summarized current research on defect accumulation under irradiation. He indicated the following points:

1. Experimental data on defect cluster density versus temperature shows that the density is several orders of magnitude higher in FCC metals (e.g. Cu) as compared to BCC metals (e.g. α-Fe).
2. The lower swelling rate is observed in BCC metals may be attributable to the lack of stabilization of defect clusters.
3. The high stacking fault energy of BCC metals allows small loops to be mobile, and hence may reduce their stability.
4. When we consider alloying elements, it may be concluded that substitutional elements block the one-dimensional migration of interstitial clusters, and hence increase their density.
5. The lack of vacancy clusters as a result of cascade restructuring events may be one reason for reduced swelling rates in BCC metals as compared to FCC types. The initial collapse of the cascade contents into a vacancy cluster in the center, and interstitial clusters on the periphery leads to the 'Production Bias' effect. Because of the small probability of cascade collapse in BCC metals, the production bias that drives swelling is small.
6. The relative roles of 'Absorption Bias' versus 'Production Bias' in driving the swelling of metallic alloys are not well established at the moment.
7. One experimental observation, which has not yet been explained, is the formation of 'rafts' of small point defect clusters in the close vicinity of dislocations. Loop rafting is thought to be linked to dislocation channeling under deformation conditions, and may thus be important in understanding flow localization problems.

11.4. Accumulation kinetics

Review of the literature indicates that production and accumulation of defects are two sides of the same problem. The detailed rate theory model of Dr Golubov (Obninsk, Russia) shows the importance of including cascade-induced clusters in the formulation, particularly for the case of neutron irradiation. Experimental evidence shows that the type of bombarding particle has a profound effect on the final swelling result. In a detailed model, both the production and the absorption bias representations of defect behavior are now included. This treatment correctly accounts for the asymmetry of point defect production and absorption, and hence may explain swelling data of FCC versus BCC metals.

11.5. Atom transport

Professor Wollenberger provided experimental information on atom transport. One adequate method for determination of the fraction of freely migrating defects (FMDs) is experimental measurements of the self-diffusion coefficient. Freely migrating defects are those who migrate randomly and thus cause diffusion and atom transport. This is somewhat of a different definition of FMDs from the definition based on escape from correlated recombination within the cascade itself. Diffusion measurements can separate the information on the sink strength on the one hand, and FMDs on the other. No one has done these experiments under neutron irradiation because of the lack of depth resolution in this case. Heavy ion irradiation provides depth resolution of about 2 nm, and can hence be sensitive to atom transport. Professor Wollenburger emphasized that experimental diffusion measurement methods should be explored, in addition to TEM observations of defect structures.

11.6. Damage accumulation at high dose

Professor Muroga (National Institute for Fusion Science, Japan) presented a summary of current research on microstructure evolution at high dose (~200 dpa). The situation at these high doses is very complex, because of the production of large concentrations of helium and transmutations, leading to drastic changes in the properties of materials. The interaction of helium and other transmutations with the microstructure is very strong, and should be considered for alloy development of fusion structures. It was emphasized that materials should be re-designed to deal with the large variations in their microstructure and chemical composition after long neutron exposures.

11.7. Effects of temperature transients

Professor Yoshida (Kyushu University, Japan) made significant points to emphasize the need to consider transient temperature effects on the evolution of irradiated microstructure. It is experimentally shown that a temperature transient can eliminate certain features of the microstructure, particularly critically stable vacancy clusters. Also, temperature transients can lead to fast migration of helium to the grain boundaries, and hence result in drastic embrittlement as a result of possibly one transient excursion. It is therefore emphasized that the interpretation of experimental data should be based on accurate monitoring of time-at-temperature, and assisted by appropriate models for their interpretation.

11.8. Mechanical properties

Dr Max Victoria (Paul Scherrer Institute, Switzerland) presented recent experimental data on one of the most significant problems in irradiation effects on the mechanical properties. He showed that, under irradiation, dislocation channels suddenly form, and plastic instabilities propagate from one end of the sample to the other. The phenomenon is observed in both FCC (i.e. Cu) and BCC (i.e. α-Fe), and is shown to be a precursor to fracture at Triple Point Junctions (TPJs). Presently, very little theoretical and experimental details are available to determine the exact nature and conditions of this phenomenon.